Effect of Laser Rescanning on the Grain Microstructure of a Selective Laser Melted Al-Mg-Zr Alloy

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Abstract

The microstructures of alloys created via Additive Manufacturing (AM) can vary substantially from those present in cast or wrought products, due to the very rapid solidification associated with AM. While numerous studies have investigated the process-microstructure relationship of alloys created by Selective Laser Melting (SLM), few have investigated the effects of laser rescanning to alter the microstructure or take advantage of the rapid solidification conditions the process provides. This study investigates the effect of single- or multiple pass laser scanning upon the grain structure of Addalloy™, a new Al-Mg-Zr alloy strengthened via L1₂ Al₃Zr precipitates. The bottom of the melt pools consisted of fine equiaxed grains (1.3 μm) that nucleated from primary Al₃Zr (100–400 nm) precipitates. The top of the melt pool consists of columnar grains (up to 40 μm long), consistent with lack of Al₃Zr nucleants due to Zr solute trapping from increased solidification velocities. Additional laser scanning (a second or third scan) reduces the amounts of columnar grains and increased the number equiaxed grains. The change is attributed to a shallower melt pool remelting the columnar grain region upon rescanning, due to reduced laser energy absorption and increased heat losses in the solid.

Keywords: Additive Manufacturing, Addalloy™, Aluminum, Rapid Solidification

1. Introduction

Additive manufacturing, also known as 3-D printing, refers to the group of technologies where a part is built up layer by layer unlike conventional manufacturing where material is removed until the desired part geometry is reached. Aside from the benefit of less waste, AM enables the rapid manufacturing of complex part geometries with minimal custom tooling. AM may further enable designs and repairs that can significantly impact device efficiency and economics, which explains the recent intense focus from industry and academia. In particular, powder-bed-based AM in the form of Electron Beam Melting (EBM) and Selective Laser Melting (SLM) has recently gained considerable attention. From a thermal history standpoint, EBM and SLM are complex manufacturing processes.[1] The melting and solidification of the powder layers are rapid with cooling rates on the order of 10^4 K/s. After the first layer, there can be tens to hundreds of additional layers each with their own thermal contribution (remelting and reheating) to the previous layers. This results in a complex thermal cycle that creates high residual stresses and non-uniform microstructures. In addition, defects such as hot cracking can occur due to the rapid solidification and reheating.[2] The complex thermomechanical history of AM leads to new processing challenges as well as new microstructure optimization opportunities.

The AM microstructure typically consists of columnar grains oriented in the build direction, but regions of equiaxed grains are also observed [1]. The grain and precipitate microstructure of any material depends on its thermo-mechanical history from solidification to heat treatment. In AM, this history is affected by numerous factors, e.g., the alloy composition, the specific AM process, and the scanning strategy. For example, studies have observed repeating patterns in the microstructure of an SLM fabricated CM247LC (a Ni superalloy) that was correlated to the scanning strategy implemented.[3] Electron beam melting (EBM) of Ni-based superalloys has resulted in in-situ precipitation [4] but superalloys fabricated with SLM have minimal to no precipitation.[5] This difference in precipitation behavior can be attributed to the high preheat temperatures in the EBM process that leads to in-situ aging. The high number of variables present in the AM process means that there are numerous ways to control the microstructure.

It is well known that grain microstructure has a major influence on materials properties. For example, a fine grain structure is beneficial for tensile and fatigue strength, but not for creep resistance. A columnar grain microstructure is more beneficial for creep resistance. For any given application, the choice of grain microstructure (columnar, equiaxed, duplex, fine or coarse, etc) can be just as important as the choice of alloy and phases. Therefore, understanding grain microstructure formation and evolution during AM is of great importance. An AM part can, in a simplistic way, be viewed as a multi-pass weld; therefore it is important to relate microstructure to the solidification (thermal gradient and velocity) conditions. A study on EBM of IN718 found agreement between observed grain microstructure and that predicted by a solidification map [6]. It is known that one can change the microstructure by controlling the thermal gradients and velocities. It was shown in IN718

that changes in local grain crystallographic orientation (transitioning from columnar to equiaxed microstructure) can be controlled by the scanning strategy [7]. One downside to this modification strategy is that some scanning strategies can result in long process times and high residual stresses and part defects. Another grain microstructure modification strategy is to add grain inoculants to the SLM powders. Martin *et al.* [8] added nanoscale hydrogen-stabilized Zr particles to the surface of Al powders to preferentially nucleate equiaxed grains during solidification. The equiaxed grain microstructure can accommodate the strain induced by the AM process, thus resulting in crack-free parts. While the strategy was successful, it requires addition of nano-size particles to the Al powder which adds to the process complexity and cost. Therefore, it is desirable to develop more than one modification strategy to allow for grain microstructure tailoring, when altering the scan strategy or the powder production are not possible. Laser rescanning is a strategy where the solidified layer is scanned again by the laser prior to the deposition of the next powder layer, thus remelting the material. Kenel *et al.* [9] showed that laser rescanning can reduce the cracking density for an oxide dispersion strengthened TiAl alloy. Yasa *et al.* [10] showed that laser rescanning on 316L stainless steel results in increased part density, better surface finish, and a finer grain microstructure. Little information exists on the potential benefit of laser rescanning on the grain microstructure of aluminum alloys, which is the focus of the present study.

Most research on AM of aluminum alloys focuses on the Al-Si system due to the castability of this alloy family [11–13]. A Si-free alloy - Al-4.6Mg-0.66Sc-0.42Zr-0.49Mn (wt.%), Scalmalloy™, developed by the Airbus Group - was shown to be processable via SLM [14]. Spierings *et al.* [15] further investigated the SLM processing of Scalmalloy and obtained densities greater than 99% with porosity as the only observed defect. A duplex grain microstructure was observed with equiaxed grains at the bottom of the melt pools and columnar grains as the remainder. Spierings *et al.* [16] attributed the fine grained regions to the inoculating effect of primary Al₃(Sc,Zr) precipitates. The columnar grain regions were attributed to the lack of Al₃(Sc,Zr) precipitation resulting from the high temperature gradients toward the melt pool surface. The duplex grain microstructure of these L1₂-precipitation-strengthened alloys makes them ideal candidates for studies on grain and precipitate microstructural control via processing parameters.

The present study investigates the effect of laser rescanning on the grain microstructure of Addalloy™, a Sc-free Al-Mg-Zr aluminum alloy (Al-3.6Mg-1.1Zr, wt.%, from NanoAl, Skokie, IL, USA) which exhibits L1₂-structured Al₃Zr nano-precipitates [17]. This alloy was specifically developed to provide, after additive manufacturing, high strength and ductility at ambient temperature as well as high creep- and coarsening resistance at elevated temperatures [17]. Aluminum alloys with less than ~0.7 wt.% Zr processed via conventional casting have been studied extensively [18–22], but little research has been conducted on Zr-rich alloys, with much higher liquidus temperatures and very broad liquidus-solidus ranges. In this paper we show that Addalloy™ can tolerate higher Zr content due to the very rapid solidification achieved via SLM, thus providing higher Zr content for meta-

stable L1₂-Al₃Zr. We study here the potential for grain microstructure modification via single and multiple laser scanning on Addalloy™ which displays a duplex grain structure and L1₂-structured Al₃Zr nano-precipitates.

2. Materials and Methods

The Addalloy™ powder utilized in this study contains 3.60 wt.% Mg and 1.12 wt.% Zr, with the balance being aluminum. The powder was produced via gas atomization by Nanoval (Berlin, Germany). The manufacturer reported a 25-45 µm size range with a D₅₀ = 37 µm.

A ConceptLaser (Concept Laser, Germany) M2 machine, with a 200 W Nd-YAG fiber laser with a 1070 nm wavelength operated in continuous wave mode, was utilized for this study. Argon shielding gas was used to keep the oxygen content in the chamber below 1% during laser processing. Sample cubes (8x8x8 mm³ and 10x10x10 mm³) were fabricated with the full 200 W laser power, a 200 mm/s laser scanning speed, a 0.135 mm hatch distance, and a 40 µm powder layer. An island scanning strategy, as described by Carter *et al.* [3], was utilized for the study of number of rescans and a unidirectional scan strategy, also described by Carter *et al.*, was utilized to study the effect of rescan orientation. A change in scanning strategy (from Chess to Unidirectional) to study rescan orientation had to be made due to machine software constraints. The study on the number of rescans was conducted so that the rescan laser path followed the same path as the original melting path. Once the rescans were completed, the powder was distributed and the island scanning strategy resumed. All other machine parameters (laser power, hatch distance, layer thickness) were kept constant. In addition, the laser rescan orientation was modified so that the rescan occurred at a 90° angle to the original melting scan. Single line scans (200 W, 200 mm/s) were performed on an aluminum base plate with a 70 µm powder layer. The solidified melt trace was rescanned with the same parameters.

Sample cubes were cold mounted in epoxy, ground, and polished with 1 µm monocrystalline diamond suspension. Final polish was done with 50 nm colloidal silica. Sections parallel to the build direction were characterized. Samples for Electron Backscatter Diffraction (EBSD) were ion-milled prior to scanning. Microstructure assessment was done with a FEI NanoSEM 230 in backscatter mode and a FEI Helios FIB in MD (Mirror detector) mode. Energy Dispersive Spectroscopy (EDX) mapping was performed with the FEI NanoSEM 230 with an Oxford Instruments detector. EBSD mapping of the samples was performed with a Tescan Lyria FIB with an EDAX EBSD detector. Scan areas were 200x150 µm at 750x magnification with a 0.2 µm step size. The TSL OIM Analysis 7 software was used to generate EBSD grain orientation maps (unique colors for grains). Equivalent grain sizes were determined based on a circular approximation based on the area of the data points in the grain. ImageJ was used to threshold SEM images for particle size analysis.

Lamellae for transmission electron microscopy (TEM) analysis were extracted from the fine grain region and coarse grain region of the single scanned sample with a FEI Helios NanoLab 600i focused

ion beam (FIB). The lamellae were extracted such that the transverse structure was showing. Since Ga ions were utilized in the thinning of the lamellas, Ga was detected on the grain boundaries and edges of the precipitates. The electron diffraction data was obtained at 200 kV with a JEOL 2200FS microscope. Scanning transmission electron microscopy (STEM) was performed on a FEI Titan Themis microscope operated at 300 kV and equipped with a probe spherical aberration corrector and a SuperEDX system (ChemiSTEM technology) with four silicon drift detectors for energy-dispersive X-ray (EDX) spectroscopy. A convergence semiangle of 25 mrad was used in combination with an annular dark field (ADF) detector with inner and outer collection semiangles of 53 and 200 mrad, respectively.

3. Results and Discussion

3.1 Single-Scan Microstructure

The density of the single-scanned sample exceeds 98% (as measured via the Archimedes method) and few defects (trapped oxides, voids) were observed on cross sections. The microstructure of the single-scanned sample is shown in Figure 1 and an EBSD Image Quality (IQ) map is shown in Figure 2. The duplex grain microstructure consists of fine-grain, equiaxed regions and coarse-grain, columnar regions. Columnar grains (1-5 μm wide and up to 40 μm in length) are oriented in the build direction, which is expected from the rapid and directional solidification process. However, these columnar grain region do not cross melt pools as observed in other alloys such as CM247LC [3]. Rather, fine grain regions are located at the bottom of the melt pools and consist of ~ 1.3 μm diameter equiaxed grains (number average). As shown in Fig. 1, sub-micron cuboidal precipitates (100 - 400 nm in size) are present in the fine-grain regions but not in the columnar regions. EBSD analysis showed that the fine-grain region is textureless and the columnar grain region displays a slight (001) texture. (Figure 2). Similar fine- and columnar grain regions, with and without primary $\text{Al}_3(\text{Zr},\text{Sc})$ micron precipitates, have been observed by Spierings et al. [15] in an Al-Mg-Sc-Zr alloy (Scalmalloy) processed by SLM.

ADF-STEM images of the fine- and coarse-grain regions are shown in Figure 3. In the lamella from the fine-grain region, precipitates are present both within the grains and at the grain boundaries. Primary Al_3Zr precipitates were not observed within the lamella from the coarse grain region. High resolution ADF-STEM images and selected area electron diffraction patterns of the cuboidal precipitates of the fine-grain region are shown in Figure 4. The atomically resolved ADF-STEM images of the precipitate-matrix interface, recorded along the [001] zone axis, show the coherent nature of the precipitates with the matrix. Additionally, the FFT pattern obtained from the precipitate (Figure 4a) clearly evidences the presence 100- and 010-type reflections which are not present in the FFT pattern of the aluminum matrix (Figure 4b). They are characteristic superlattice reflections of the L_{12} phase,

and are also visible in the selected area diffraction pattern obtained from a second precipitate (Figure 4c).

STEM-EDX chemical mapping of the fine grain region and coarse grain region is shown in Figure 5. As illustrated in Figure 5a, the cuboidal precipitate highlighted in Figure 3a is enriched in Zr. Since the precipitates are identified as having the $L1_2$ structure, it can be concluded that they are $L1_2$ -structured Al_3Zr precipitates. Also, a Mg- and O-rich region is observed in the precipitate, suggesting the presence of an oxide inclusion, about ~25 nm in size. The Zr-depleted region in the center of precipitate is believed to be due to porosity but the origin of the porosity is unknown.

The coarse-grain region contains numerous spheroidal particles (~80 nm) enriched in Mg and O that are likely oxides originating from the powder surface, possibly dissolved and reprecipitated in the melt pool, given their spherical shape. Fe-rich precipitates (~30 nm) are also present at the grain boundaries and are attributed to Fe-impurities in the powder. A uniform distribution of Zr (up to 0.5 at% as evidenced by EDX) was observed within the mapped area and no areas were suggestive of Zr precipitates (cf. also Figure S1 in the supplementary file).

Considering the Al-rich part of the Al-Zr phase diagram shown in Figure 6 [23], it is expected that, in Fig. 1, the particles are primary $L1_2$ -structured Al_3Zr primary precipitates formed in the liquid during solidification. This is in accordance with previous work in which the formation of metastable $L1_2$ -structured Al_3Zr in rapidly solidified Al-Zr alloys was reported [24–26]. The dashed line in Figure 6 shows the metastable $L1_2$ solvus in the solid state which was determined employing ab initio calculations by Liu et al. [27]. Smaller nm-scale particles (white in the backscatter electron imaging) are observed throughout the structure. SEM EDX analysis of the particles was not possible due to their sub-250 nm size and they were not observed in the TEM foils prepared, but they are likely Al_3Zr secondary precipitates formed in the solid alloy during the repeated heating inherent to SLM. Such $L1_2$ -structured precipitates form in cast Al-Zr binary alloys with lower Zr content (and thus despite a lower precipitation driving force) upon aging in the temperature range 375–600 °C [18,19].

The duplex grain microstructure and inhomogeneous precipitate distribution may be explained by the change of the solidification front velocity at different regions of the melt pool and the order in which the phases form, as schematically shown in Figure 7. The Al-rich end of the Al-Zr binary phase diagram displays a peritectic reaction (cf. Figure 6) [23][28], meaning that the solute-rich phase, Al_3Zr (assumed to be in the metastable $L1_2$ structure) is the first to form in the melt on cooling. These primary Al_3Zr precipitates then serve as inoculants for the nucleation of fine, micron-size fcc-Al grains, as discussed by Knipling *et al.* [28]. The absence of equiaxed grains toward the surface of the melt pool is consistent with the suppression of this inoculation effect due to an increase in the solidification front velocity. In laser-based processing, it is known that the solidification front velocity increases, approaching the beam scanning velocity as the melt pool surface is approached [29]. For example, pronounced microstructure changes over the depth of the melt pools (i.e., eutectic growth at the bottom and dendritic/cellular growth at the melt centerline) have been observed for a SLM processed Al-Ce eu-

tectic alloy and were attributed to solidification front velocity increases [30]. In our alloy, it is expected that the solidification front velocity increased near the pool surface to the point that Zr was trapped in the solidifying fcc-Al grains, remaining in supersaturated solution. SEM and TEM EDX measurements taken in the columnar grain region of numerous melt pools showed an average Zr concentration of 1.5 wt.%, which is much higher than the equilibrium maximum solid solubility of 0.28 wt% Zr for stable $D0_{23}$ - Al_3Zr or even the 0.80 wt.% value for metastable $L1_2$ - Al_3Zr in binary Al-Zr [28]. Since Zr is present in the columnar grain regions but primary Al_3Zr precipitates are not, solidification velocities are high enough to suppress Zr partitioning (cf. Figure 3 and Figure 5). Knipling et al. [25] summarized the cooling rates required to suppress the formation of Al_3Zr and Al_3Ti in binary Al-Zr and Al-Ti alloys, respectively. According to their work, a cooling rate between 10^2 and 10^3 Ks^{-1} is sufficient to prevent the formation of Al_3Zr . These cooling rates can be easily achieved in the melt pools during SLM processing. With no primary Al_3Zr present due to solute trapping, inoculation is inoperative close to the surface of the melt pool, consistent with the change to epitaxial columnar grain growth oriented parallel to the direction of the heat flow. However, the totality of the Zr is available for precipitation of nanosize Al_3Zr precipitates upon subsequent aging in the columnar grains, with the potential to make them stronger than the equiaxed grain where less Zr is present in solid solution.

The above scenario for our Al-Zr-based alloy is similar to that proposed by Spierings *et al.* [16] for their Al-Sc-Zr-based alloy, who also identified primary $Al_3(Sc,Zr)$ precipitates as nucleant for their fine-grain Al regions deep in the melt pool, and the suppression of primary precipitation as the reason for the columnar grain regions near the melt pool surface. However, given that the Al-Sc and Al-Zr systems are eutectic and peritectic, respectively, the primary precipitation of mixed $Al_3(Sc,Zr)$ precipitates is more complex than in the present simpler Al-Zr peritectic system.

Figure 1: (a,b) BSE images of the single-scanned sample showing a duplex grain microstructure consisting of regions with columnar grains and equiaxed micron-size grains (each containing one primary Al_3Zr precipitate).

Figure 2: (a) EBSD Image Quality (IQ) map of single-scanned specimen showing the columnar and equiaxed grain regions. A melt pool boundary is shown by a black line. (b) EBSD pole figure from the fine grain region marked b in the IQ map. (c) EBSD pole figure from the coarse grain region marked c in IQ map.

Figure 3: (a) Low magnification ADF-STEM image taken from the fine-grain region showing a primary Al_3Zr L_{12} precipitate within a micron-size grain. (b) Low magnification ADF-STEM image taken from the coarse-grain region showing precipitates within grains and at grain boundaries. Dotted lines indicate the locations of the EDX maps shown in Figure 5 where these precipitates are identified.

Figure 4: (a) High resolution ADF-STEM image, along the $[001]$ zone axis, of the upper left corner of the precipitate in Figure 3 and corresponding FFT of the highlighted precipitate area (black-rimmed square). (b) High resolution ADF-STEM image of the lower left corner of the precipitate in Figure 3 and corresponding FFT of the highlighted matrix area (white-rimmed square). (c) Selected area electron diffraction pattern along $[001]$ taken from a separate precipitate showing the 100- and 010-type superlattice reflections of the $L1_2$ structure.

Figure 5: (a) STEM-EDX map of the precipitate shown in Figure 3. The precipitate is enriched in Zr and contains a region enriched in Mg and O consistent with a fine oxide particle. (b) STEM-EDX map of the coarse grain region. Zr was detected at 0.5at% uniformly throughout the sample. Numerous spherical particles enriched in Mg and O are present, and Fe-rich precipitates are present at the grain

boundaries. A uniform distribution of Si was also observed but is not shown in the figure. Iron-rich precipitates present at the grain boundaries are believed to be impurities. Chemical analysis was performed using the Zr-K, Al-K, Mg-K, Fe-K and O-K lines.

Figure 6: Equilibrium Al-rich Al-Zr binary phase diagram (adapted from Murray [23]) with metastable Al_3Zr (L_{12}). [25]

Figure 7: Schematic of melt pool and associated plot of solidification front velocity as a function of depth z in the melt pool. Solidification front velocity increases towards the surface of the melt pool, crossing a threshold of solute trapping preventing the precipitation of Al_3Zr precipitates.

3.2 Multi-Pass Scan Microstructure

Optical micrographs of double- and single-scan samples are shown in Figure 8. The melt pools exhibit a reduced depth with laser rescanning; using the line intercept method, the single- and double-scan samples show 10 and 14 pool boundaries over a 500 μm length, respectively. Archimedes density of the samples did not increase with additional scans. EBSD grain maps showing the effect of repeated laser scans for each layer are shown in Figure 9a. Area fractions of the equivalent grain sizes are shown in Figure 9f which indicates that additional laser scanning refines the grains: each rescan reduces the area fraction of the coarser, elongated grains and increases the area fraction of the finer, equiaxed grains.

The shallower melt pools and subsequent grain refinement observed in double- and triple-scanned samples are likely due to the difference in laser-material interaction and heat transfer conditions. Figure 10 shows a schematic of the melt pool structures, with an image from a double-scanned single line track showing the melt pool depth difference (30 μm surface penetration from first scan with powder layer; approximately 5 μm penetration from second scan on bulk surface). Thus, the shallower melt pool for the second scan does not completely remelt the first melt pool but rather remelts preferentially the columnar grain structure near the top of the pool, while the equiaxed grains at the bottom of the first pool remain unmelted. It can be assumed that the solidification in the second shallower melt pools occurs in the same way as in the original melt pools, with the formation of equiaxed grains nucleated by primary Al_3Zr precipitates in the lower part of the pool and elongated, Al_3Zr -free grains near the top, when the high solidification front velocity inhibits the formation of the Al_3Zr inoculating particles. The new equiaxed grains form in place of the columnar grains created during the first scan. It is important to note that the melt pool of the first scan (melting the powder) must not be deep enough to completely remelt the previously solidified layer; otherwise the rescan-influenced microstructure will be destroyed upon melting of the subsequent powder layer.

Many factors can influence the melt pool geometry and thereby the solidification conditions. An important contributing factor to a deeper melt pool on the first scan is related to laser-material interaction. Surface roughness is known to have a major influence on laser absorptivity in metals [31,32]. Tolochko *et al.* [33] reported that the 1.06 μm wavelength laser absorptivity of W and Pb powders is nearly twice that of the consolidated material, due to the scattering in the gaps between powder particles. It can be assumed that this is also the case for the present Al alloys. A reduction in the absorptivity from powder to consolidated material would result in reduced energy transfer during the second scan, and thus remelting of a shallower, narrower material volume. The influence of the surface roughness (a smoother surface has higher reflectivity) on the melt pool size and geometry for laser scanned Addalloy is qualitatively demonstrated in Figure S2 in the supplementary file. Even though the experiment was performed without powder on the sample surface, it is observed that for the same laser parameters a rough surface leads to the formation of a wider and deeper melt pool than a smooth surface.

It is further known that the thermal conductivity of the base material in laser welding influences the depth of the melt pool [34] and that absorptivity losses can reduce the energy input from the laser [35]. For a SLM process when each layer is scanned only once, the laser radiation and resulting melt pool interact with a layer of powder (Fig. 10). In the case of rescanning, the laser radiation and subsequent melt pool interacts with a powder layer (on at least one side of the pool) in the first scan, but upon remelting in the second scan, the pool is surrounded with dense material, which is changing the heat transfer conditions. Since powders are less heat-conducting than dense material, heat transfer from the melt pool into the powder direction may be reduced as compared to directions where solid material is in contact. With laser rescanning of the already consolidated part, heat can dissipate more uniformly in all directions. The reduced heat transfer of the powder-contacting melt could then lead to a more pronounced heat transfer to the dense alloy below and thus to a deeper melt pool.

Several studies indicate that, during laser-based processing, the beam mostly interacts with the liquid metal [36][31]. Laser-liquid interaction may also vary between the single-scan and rescan cases. Matthews *et al.* observed that the melt pool during powder based processing is very dynamic [37]. Kaplan *et al.* showed that the absorptivity of the molten weld pool can change based on the surface conditions of the melt [36]. A wavier surface was shown to improve the absorptivity for materials processed with a 1 μm wavelength laser. It is conceivable that the highly dynamic melt pool observed in SLM can result in surface-induced absorptivity increases over a typical laser weld melt pool, as would be the case with the rescanning. However, experimental evidence does not exist to support or invalidate this hypothesis.

Assuming a laser interaction with the solid rather than the liquid (in the case of poor laser-material coupling), the further grain refinement resulting from the third laser pass may be explained by a further decrease in the energy input. It has been shown by Yasa *et al.* [10] that laser rescanning results in a smoother surface finish. The improved surface finish after the second scan may decrease the surface roughness enough that the absorptivity further decreases. A further decrease would result in even shallower melt pools and a further replacement of the columnar grains with equiaxed grains *via* remelting and resolidification. However, no noticeable difference of the melt pool size between the double- and triple-scanned samples was observed in metallographic cross sections. Additional heating from repeated laser scans can also influence the grain microstructure. A higher bulk temperature will result in lower cooling rates; as a result, the melt pool might have lower solidification velocities, leading to a later transition from fine- to coarse-grain solidification microstructure. It might also be expected that the additional thermal input caused by the rescanning results in some grain recrystallization; however, grains are likely to be pinned by the Al_3Zr and Fe-rich precipitates and the oxide particles.

The unidirectional scan with 90° rescan yielded an even more refined structure than the chessboard strategy with same trace rescan sample, as shown in Figure 8. However, a direct comparison be-

tween rescanning on the same laser trace and at 90° to the first laser trace for the chess-scanning strategy was not possible due to machine software limitations. The resulting crisscross of melt pools may serve to remelt the columnar grains more than rescanning on the same melt trace does. Further investigation into the impact of different re-scanning strategies is required.

This study only examined the effect of number of rescans and orientation, but further processing parameters can be changed during re-scanning. For example, the melt pool dimensions can be controlled by altering laser power and scan velocity of the rescans. It is also possible to modify the microstructure, e.g. by recrystallization or in-process aging, by adjusting the laser rescanning parameters so that the former melt pool remains in the solid state. A systematic study of the influence of different rescanning parameters and strategies on the microstructure formation in AddalloyTM and other alloys will be presented in the future.

Figure 8: Optical micrographs comparing the melt pools of (a) single-scan and (b) double-scan sample. The rescanning results in shallower melt pools.

Figure 9: a.) EBSD grain maps of (a) single, (b) double, and (c) triple scans showing grain refinement with additional scans. White areas on the EBSD grain maps are regions filtered out; d) Distribution of area fractions vs. grain diameter, showing that rescanning reduces the fraction of large columnar grains and increases the fraction of fine, equiaxed grains

Figure 10: Schematic of melt pool depth differences between single- and double-scan cases. Reduced laser absorptivity on a consolidated surface and increased heat transfer result in a shallower melt pool during the rescan. Scan direction is perpendicular to image plane. The micrograph of a double-scanned single line shows the melt pool boundaries (arrows).

Figure 11: EBSD grain maps comparing double-scans where the second scan is performed (a) parallel or (b) perpendicular to the first scan. White areas on the EBSD grain maps are re-gions of low confidence index (0.1) (c) Area fractions of each grain diameter showing that rescanning perpendicular to the first scan results in more grain refinement than rescanning in a parallel direction.

3. Conclusions

The effect of repeated laser scanning during SLM on the grain microstructure of a precipitation-hardening Al-Mg-Zr alloy (Addalloy™) was studied. The as-processed alloy (single scan) has a duplex grain microstructure consisting of regions of fine, equiaxed grains (1.3 μm in diameter) at the bottom of the melt pools and regions of coarse, columnar grains ($\sim 40 \mu\text{m}$ long) at the top of the pool. Sub-micron (100-400 nm), cuboidal primary Al_3Zr precipitates act as nucleation sites for the equiaxed fcc-Al grains. No primary precipitates are visible in the columnar grain regions, consistent with the lack of nucleation: this is attributed to Zr solute trapping from increasing solidification front velocities as the melt pool surface is approached.

Laser rescanning of the consolidated Al-Mg-Zr alloy results in grain refinement which is attributed to the formation of shallower melt pools upon rescanning due to reduction of laser energy transfer to the solid and modification in heat transfer in the solid. The shallower melt pool upon rescanning melts the columnar grain region at the top of the original melt pool but not the lower equiaxed region.

The equiaxed grains from the original scan remain and the columnar region is replaced with new equiaxed grains (and some new columnar grains) from the new melt pool. Further grain size reduction is observed when the alloy is subjected to a third rescanning. Laser rescanning is an easily implemented option for in-process microstructure modification, albeit at the cost of longer processing times.

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